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Some microstructural and alloying effects upon low cycle fatigue life of pressure vessel steels

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SOME MICROSTRUCTURAL AND ALLOYING EFFECTS UPON LOW CYCLE FATIGUE LIFE OF PRESSURE VESSEL STEELS

by

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CERTIFICATE OF APPROVAL

This thesis is accepted and approved in partial fulfillment of the requirements for the degree of Master of Science.

September 19, 1963
(date)

Alan W. Pense
Professor in Charge

[illegible signature]
Head of the Department

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ABSTRACT

Fatigue is one of several basic modes of failure and is of primary consideration in the case of pressure vessel service. Since pressure vessels exhibit an average service life of up to 100,000 cycles, the strains involved necessarily result in a degree of plastic deformation. For this reason, the low cycle fatigue life of pressure vessel type steels, as affected by microstructural changes and alloy content, is of interest to material suppliers and fabricators.

The Lehigh cantilever fatigue specimen was utilized for the study which encompassed the 5,000 to 100,000 cycle range. Specimen loading was accomplished by a constant deflection reversed bending technique at a rate of 200 cycles per minute. Strain measurements were made using a Tuckerman optical strain gage.

The microstructural effect was observed by comparing the fatigue resistance of specimens of equal strength level but different heat treatment histories. In all cases the normalized structure proved to be superior to the corresponding spray quenched and tempered condition. Where similar structures were compared, the fatigue life was found to be closely related to tensile strength for 100,000 cycle life. The low alloy content steels revealed a marked response of heat treatment upon fatigue resistance while the plain carbon steel showed a negligible effect. Increased alloy content altered the resulting slope of the fatigue curves and effected a grouping of the curves as to material composition. Strain magnitudes and alloy composition also had noticeable effects upon crack propagation rates in low cycle fatigue.

INTRODUCTION

A good deal of work has been performed in the past by investigators studying the mechanical properties of pressure vessel type steels. Within the past ten years much of this work has become centered about fatigue resistance. While few pressure vessel failures can be attributed directly to a fatigue-type of failure, one must not overlook the possibility of a crack, initiated by fatigue and going unnoticed and unrepaired, leading to a final cataclysmic rupture at some future time. This thought has concentrated studies upon the effect of design and notches on fatigue life and on relationships of fatigue resistance to other mechanical properties. However, in spite of this general interest there is still a scarcity of information regarding microstructural and alloying effects upon low cycle fatigue life.

Among the few studies of microstructure that have been undertaken Cagaud[1] reports that the type of structure formed by quenching followed by a tempering operation was found to be desirable for good fatigue resistance. While the general quenched and tempered structure was the most promising from an endurance standpoint, a tempering temperature of 650 to 850°F which provided a fatigue limit of maximum value also resulted in minimum impact strength. Such a microstructure cannot be utilized in many applications and especially pressure vessel service which requires a better balance of these two properties. This leads to higher tempering temperatures in an attempt to bring the impact and fatigue properties into a better relationship.

Cagaud also found that unless there is a unique circumstance requiring special properties, the utilization of high strength steels is not necessary for adequate fatigue resistance in most cases.

Dolan and Yen(2), and Sinclair and Dolan(3) found the quenched and tempered martensitic structure to have higher fatigue strength than a pearlite-ferrite structure of the same hardness value. These tests were performed on both notched and un-notched rotating beam type specimens and therefore consider not only the effects of microstructure but also the effect of notch-sensitivity upon endurance limit.

Kooistra(4) reports that composition, microstructure, and previous mechanical work influence fatigue resistance in the same manner as they affect tensile strength when relatively small strains are being considered. There is no apparent fatigue resistance indicator or relationship between fatigue and other mechanical properties where high strains are applied.

Gross, Gucer, and Stout(5) state that heat treatment apparently affects the fatigue resistance of pressure vessel steels only when the tensile strength is also considerably affected. Their work revealed a direct relationship existed between tensile strength and fatigue life.

Little, if any, work has been reported which compares low cycle fatigue resistance with the corresponding alloy contents of the steels tested.

The purpose of this study was to establish the relationship between microstructures produced by commercially feasible heat treatments and room temperature fatigue resistance in pressure vessel steels. The specimen used was chosen to closely approach pressure vessel service conditions in both size and type of loading.

While most fatigue work deals with an endurance limit running over one-million cycles, the average life of a pressure vessel is normally considered to be up to 100,000 cycles. A fatigue life of 5,000 to 100,000 cycles, therefore, was the range of interest in this study.

The strains involved in the 5,000 to 100,000 cycle range differ from those normally considered in fatigue work dealing with endurance limits in that they are larger and generally are considered plastic strains. By plastic strains it is meant that they are beyond the elastic limit of the material. The behavior of steel as governed by microstructure under these higher strains can differ quite markedly from that when considering more than 1,000,000 cycle life.

The reported study is not interested in the endurance limit since primary interest lies in material behavior over this lower range of fatigue life.

EXPERIMENTAL PROCEDURE

Steels and Heat Treatments

The steels of interest in the study were A-302; A-212, fine-grain; A-212, coarse-grain; HY-80; and A-387, grade B. The chemical analyses of these materials can be found in Table I. These steels were tested in the following conditions of heat treatment:

1. A-302, Al killed
   a) Normalized from 1650°F
   b) Spray quenched from 1650°F, tempered 1 hr. at 1300°F
2. A-212, Al killed
   a) Normalized from 1650°F
   b) Spray quenched from 1650°F, tempered 1 hr. at 1300°F
   c) Spray quenched from 1650°F
   d) Furnace cooled from 1650°F
   e) Plate cooled from 1650°F
3. HY-80
   a) Normalized from 1600°F
   b) Spray quenched from 1600°F, tempered 1 1/2 hrs. at 1150°F
   c) Spray quenched from 1600°F, tempered 1 1/2 hrs. at 850°F
   d) Spray quenched from 1600°F, tempered 5 1/2 hrs. at 1200°F

4. A-212, coarse grain

   a) Normalized from 1650°F

   b) Spray quenched from 1650°F, tempered 15 hrs. at 1300°F

5. A-387, grade B, coarse grain

   a) Normalized from 1675°F

   b) Spray quenched from 1675°F, tempered 2 hrs. at 1375°F

The purpose of these treatments was to enable one to study the effect of differing microstructures of like hardness. This aim was accomplished by treatments (a) and (b) of (1) through (5) for A-302, A-212 F.G., HY-80, A-212 C.G., and A-387 respectively.

In addition to this principal phase of work, various supplementary studies were also made. Treatments (c), (d), and (e) under (2) permitted observation of the effects of a range of cooling rates on a plain carbon steel (A-212 in this case). Mid-section cooling rate studies were made over the range of treatments applied to the A-212 steel; these results are found in Table II.

Treatments (c) and (d) of (3) enabled one to observe the effect of tempering at various temperatures and times upon the quenched structure of HY-80 and their subsequent effect upon fatigue life.

A comparison of A-302, A-212, HY-80, and A-387 should show any existing effect of alloy composition upon fatigue resistance while it was hoped that the effects of deoxidation practice upon fatigue life could be compared directly between the two heats of A-212.

Testing Apparatus

The fatigue testing was performed using the standard Lehigh cantilever fatigue specimen and a constant deflection type of loading. This specimen is shown in Figure 1. The specimen geometry is such that a 2:1 biaxial stress distribution is achieved. A complete description of the test has been published elsewhere[6,7]. In accordance with standard procedure a cycle rate of 200 cycles per minute was utilized. All tests were performed at ambient temperatures. The steels tested were obtained as 3/4-inch thick plate with the longitudinal axis of the specimen cut from the plate in the same orientation as the rolling direction following heat treatment.

The reduced section of the specimen was milled and subsequently ground to an 80-grit finish with the scratches running parallel to the longitudinal axis of the specimen.

Strain Measurements

Strain readings were measured on each specimen by use of a Tuckerman optical strain gage with a nominal gage length of 1/4-inch. The gage length straddled the midpoint of the reduced section and registered total strain over several complete cycles of reversed bending. Strain measurements were made and recorded after the tenth cycle to eliminate any possible errors resulting from strain hardening effects[5]. Vertical deflection readings of the specimen at the point of loading were also recorded with the strain readings.

Testing Criteria

Various total strain levels were used to obtain fatigue failure over the range of 5,000 to 100,000 cycles. Twelve fatigue specimens were used to establish each fatigue curve. All specimens were run to final failure which was defined as the point at which the flexed section could no longer transmit a useful load to the fixed end. This point was determined by zero deflection as registered by a dial gage mounted between the fracture zone and the stationary clamp. Supplementary cycle readings were also recorded when the crack depth reached 3/16-inch.

Microstructural studies were performed with respect to basic structure and crack propagation. The tensile properties were evaluated for each steel and condition using the standard ASTM .252-inch test bar and can be found in the mechanical property listings of Table III.

PRESENTATION AND DISCUSSION OF RESULTS

Figures 2 through 6 show the effect of microstructure upon 5,000 to 100,000 cycle fatigue life for A-212 F.G., HY-80, A-302, A-212 C.G., and A-387 respectively. While the effect is not as marked for the fine-grain A-212, all the steels show a general superiority of the normalized microstructure over the quenched and tempered (to a comparable tensile strength level) structure. In every case the quenched and tempered structures were more acicular than the normalized microstructures. (See Appendix A.) A metallographic crack study revealed a somewhat different crack path in the two types of structures. The more massive ferrite-pearlite mixtures of the normalized structures presented a potential crack with a widely spaced, discontinuous pathway. Since the crack favored pearlite areas, it had to traverse the larger, more ductile ferrite grains by necessity in its propagation. The acicular structure, however, permitted the crack to follow the more closely spaced carbide particles which were somewhat aligned. This carbide alignment reduced the size of the ferrite areas confronting a crack, thus leading to a semi-continuous crack path of oriented fine carbides.

This superiority of the normalized structure is apparently a result of the less continuous path presented for crack propagation, while the aligned carbides of the quenched and tempered structures actually aid propagation of fatigue cracks by providing them with an oriented acicular pathway for growth. The improvement of the normalized condition over the equal-strength quenched and tempered condition implies an effect of microstructure upon fatigue resistance without a similar effect on tensile strength.

Figure 2 shows the almost negligible effect of varying cooling rates over the range of 0.032 to 20.0 degrees per second upon fatigue resistance of A-212. This lack of response can be directly attributed to the low hardenability of the steel. The microstructures were, in all cases, primarily ferrite in ranging from an equiaxed large-grained ferrite matrix with coarse pearlite colonies to a fine acicular ferrite and unresolved pearlite mixture. The continuous ferrite matrix was the dominant constituent, however, with the only basic difference being in the spray quenched structure which was acicular in nature. This acicular structure exhibited somewhat better fatigue resistance at the 100,000 cycle level but was markedly stronger than any of the other A-212 conditions.

Figure 3 is a corresponding plot of the HY-80 group of four differently heat treated conditions. Here it can be seen that a change in strength level for similar microstructures produces a marked effect upon fatigue life throughout the 5,000 to 100,000 cycle life. For this material the resulting structures were primarily tempered martensite, the only exception being the normalized series. Here, as in the previous examples, the normalized condition exhibited better fatigue resistance than the equal strength specimens which were quenched and tempered at 1150°F for 1 1/2 hours. The specimens that were quenched and tempered at 1200°F for 5 1/2 hours had a lower tensile strength than the two previous conditions and also poorer fatigue resistance over the entire 5,000 to 100,000 cycle range. The HY-80 specimens showing the best fatigue life also had the highest tensile strength. These were quenched and tempered at 850°F for 1 1/2 hours.

The two plots of Figure 7 summarize the results of comparing the normalized with the spray quenched and tempered structures over a range of tensile strengths. The 100,000 cycle life plot clearly illustrates the increase in fatigue resistance of the normalized structure over its quenched and tempered counterpart. A definite trend can be observed to exist over the range of the comparison in that for the 100,000 cycle plot, the normalized and the quenched and tempered conditions fall on two separate parallel lines when plotted against tensile strength. These curves indicate that for a given type of microstructure, the 100,000 cycle allowable total strain is directly proportional to tensile strength. While the 5,000 cycle plot reveals this same difference between the two types of structures of equal tensile strength, no clear-cut relationship with tensile strength is evidenced.

Figure 8 compares two heats of A-212 of differing deoxidation practices. The plot shows no clear-cut advantage existed for one deoxidation practice over the other in terms of fatigue resistance. While the coarse-grained material showed poorer fatigue resistance in the 5,000 cycle level it did exhibit slightly better resistance in the low strain portion of the range than did the fine-grained steel. One must be careful, however, in reaching a conclusion concerning killing practice since the coarse-grained heat also differed from the fine-grained heat in its residual alloy content. The decrease in slope of the coarse-grained fatigue curve may also be indicative of the effect an increase in alloy content has on fatigue resistance over the 5,000 to 100,000 cycle life. This alloy effect can be seen by comparing Figure 2 (the plain carbon A-212) with any of the alloy steels represented by Figures 3, 4, or 6. The change in slope is also substantiated by previous work performed at Lehigh. In fact, by classifying the steels on the basis of plain carbon as opposed to low alloy compositions, the trend is for two separate groupings of low cycle fatigue curves; the plain carbon class showing somewhat better fatigue resistance in the 5,000 cycle end of the range and the alloy group excelling at the 100,000 cycle end with the intersection point being around the 10,000 to 20,000 cycle values[8]. With this in mind, one cannot directly relate the difference in fatigue resistance between the two A-212 heats to any one cause, and especially not to the effect of deoxidation practice.

Figure 9 illustrates a form of the aforementioned separate grouping based upon alloy content. The A-212 (fine grain) points fall on a distinct line apart from the data points of the low alloy bearing steels with the values for the A-212 (coarse grain) plus high residual content steel falling in between the two lines. This plot shows the apparent relationship between the average per cent of total fatigue life for a crack to grow to a depth of 3/16-inch and yield strength. In essence the plot indicates the relative importance of crack initiation and propagation rates as compared with alloy content. It should be emphasized that the percentage figures are average values over the complete series of specimens in question. Table IV is a listing of these percentage figures for the 5,000 and 100,000 cycle points for the individual steels and conditions. A close examination of these data reveals the interesting observation that crack initiation becomes an important aspect to consider. In the low cycle region (5,000 cycles) one would expect a given crack to grow faster than a crack involved in the 100,000 cycle life simply because of the much greater strain applied to the specimen. If this is true, one would also expect the more highly strained crack to grow to a given depth sooner with respect to total life than the crack under lower strain. This is substantiated by the data of Table IV. If the crack propagation rate is much faster for the more highly strained specimens the crack will probably initiate at a later relative time in the total fatigue life for higher strains than for low strains. The same general type of statement could be made with respect to composition on the basis of Figure 9. Here it is indicated that for a given yield strength, a crack will probably initiate and will propagate faster for the low alloy steels than it will for the plain carbon steel. These statements are based upon observations made during the later portion of crack propagation in the specimens tested. It is therefore impossible to make exact statements concerning the point of initiation of fatigue cracks in these materials.

CONCLUSIONS

1. Normalizing produces a structure exhibiting low cycle fatigue resistance which is definitely superior to that resulting from quenching and tempering to the same hardness level.
2. Accelerated cooling of plain carbon steel (A-212) through cooling rates ranging from 0.032 to 20.0 degrees per second alters the fatigue resistance over the 5,000 to 100,000 cycle life region only slightly.
3. High strength steels do not necessarily provide the optimum fatigue resistance over the entire 5,000 to 100,000 cycle lifetime span.
4. In general, allowable total strain for 100,000 cycle life is closely related to tensile strength, while such a relationship does not exist for high strains in the 5,000 cycle region.
5. Increased alloy content tends to reduce the slope of the fatigue curve over the 5,000 to 100,000 cycle region such that plain carbon steels tend to be superior in the 5,000 cycle range while the low alloy steels excel above the 20,000 cycle region.
6. Alloy content and applied strain magnitudes have noticeable affects upon crack propagation rates in low cycle fatigue.

TABLE I

THE COMPOSITIONS OF THE STEELS TESTED

Steel	C	Mn	P	S	Si	Mo	Ni	Cr
A-212 F.G.	0.29	0.73	0.013	0.024	0.21	-	-	-
A-212 C.G.	0.28	0.82	0.020	0.036	0.24	0.05	0.16	0.19
A-302	0.19	1.32	0.022	0.024	0.25	0.43	-	-
A-387	0.12	0.56	0.010	0.037	0.20	0.46	-	1.09
HY-80	0.18	0.24	0.010	0.015	0.21	0.25	2.26	1.20

TABLE II

COOLING RATES*

(Mid-section of 3/4" thick plate)

Treatment	Cooling Rate - F°/sec.
Austenitized at 1650°F	
Spray Quenching	20
Plate Cooling	2.02
Air Cooling	1.01
Furnace Cooling	0.032

*Half-Temperature-Time Cooling Rates

TABLE III

MECHANICAL PROPERTIES

Steel	Condition	Y. S. (ksi)	T. S. (ksi)	YS/TS (%)	R.A. (%)	Elong. (%)	BHN	Str.Hard. Exponent	Total Strain (%) 5,000 Cycles	Total Strain (%) 100,000 Cycles
A-302	Norm	54.3	87.8	61.8	60.6	26.0	191	0.168	1.02	0.48
Aust.-1650°F	Q.& T.1 hr. 1300°F	72.5	87.2	83.2	69.7	27.0	194	0.174	0.90	0.42
A-212 F.G.	Norm	50.6	77.8	65.2	60.4	28.0	149	0.218	1.35	0.37
Aust.-1650°F	Q.& T. 1 hr. 1300°F	51.6	77.6	67.0	71.6	32.5	146	0.191	1.30	0.36
	S.Q.	68.4	101.0	67.8	69.3	24.5	180	0.142	1.25	0.42
	A.	35.4	70.6	50.2	72.7	31.5	132	0.235	1.34	0.37
	P.C.	53.1	80.0	66.4	65.0	29.0	158	0.198	1.30	0.37
A-212 C.G.	Norm	49.4	83.6	59.0	56.7	27.0	156	0.174	1.21	0.45
Aust.-1650°F	Q.& T. 15 hrs. 1300°F	57.6	83.2	69.2	71.8	28.0	156	0.192	1.08	0.39
HY-80	Norm	80.2	125.1	64.2	60.0	21.5	229	0.129	1.10	0.57
Aust.-1600°F	Q.& T. 1 1/2 hrs. 1150°F	112.5	125.0	90.0	74.6	21.5	234	0.087	1.04	0.50
	Q.& T. 1 1/2 hrs. 850°F	156.0	166.0	94.0	62.2	16.0	332	0.091	1.14	0.67
	Q.& T. 5 1/2 hrs. 1200°F	92.3	106.0	87.0	76.6	24.0	207	0.110	0.99	0.44
A-387	Norm	39.2	77.2	51.4	61.6	26.5	151	0.166	1.10	0.46
Aust.-1675°F	Q.& T.2 hrs. 1375°F	62.5	80.7	77.5	74.3	28.0	163	0.147	1.02	0.43

KEY

Norm - Normalized
Q.& T. - Spray Quenched and Tempered
S.Q. - As Spray Quenched
A. - Annealed
P.C. - Plate Cooled

TABLE IV

CRACK INITIATION AND PROPAGATION DATA

Steel	Condition	Per cent of Total Fatigue Life for 3/16" Depth of Crack 5,000 Cycles	100,000 Cycles
A-302	Norm	89.1	84.2
	Q & T	91.8	79.6
A-212	Norm	89.6	79.9
F.G.	Q & T	87.6	77.4
	S. Q.	89.4	62.4
	A	89.5	81.3
	P. C.	89.9	74.0
HY-80	Norm	93.5	81.5
	Q & T-1150	89.5	73.4
	Q & T-850	91.9	67.3
	Q & T-1200	89.5	76.3
A-212	Norm	93.5	78.0
C.G.	Q & T	92.4	79.8
A-387	Norm	94.6	79.5
	Q & T	92.1	89.8

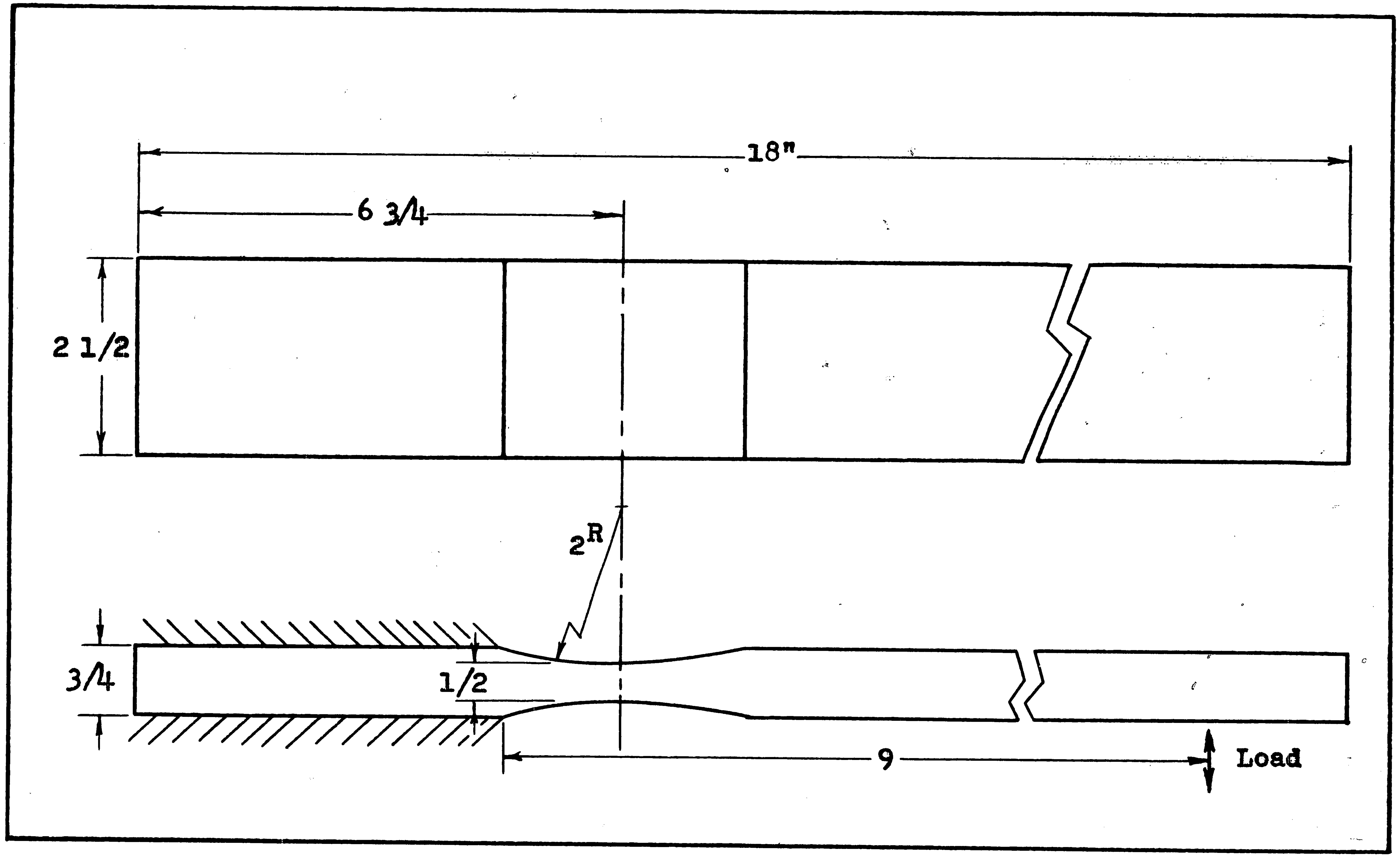
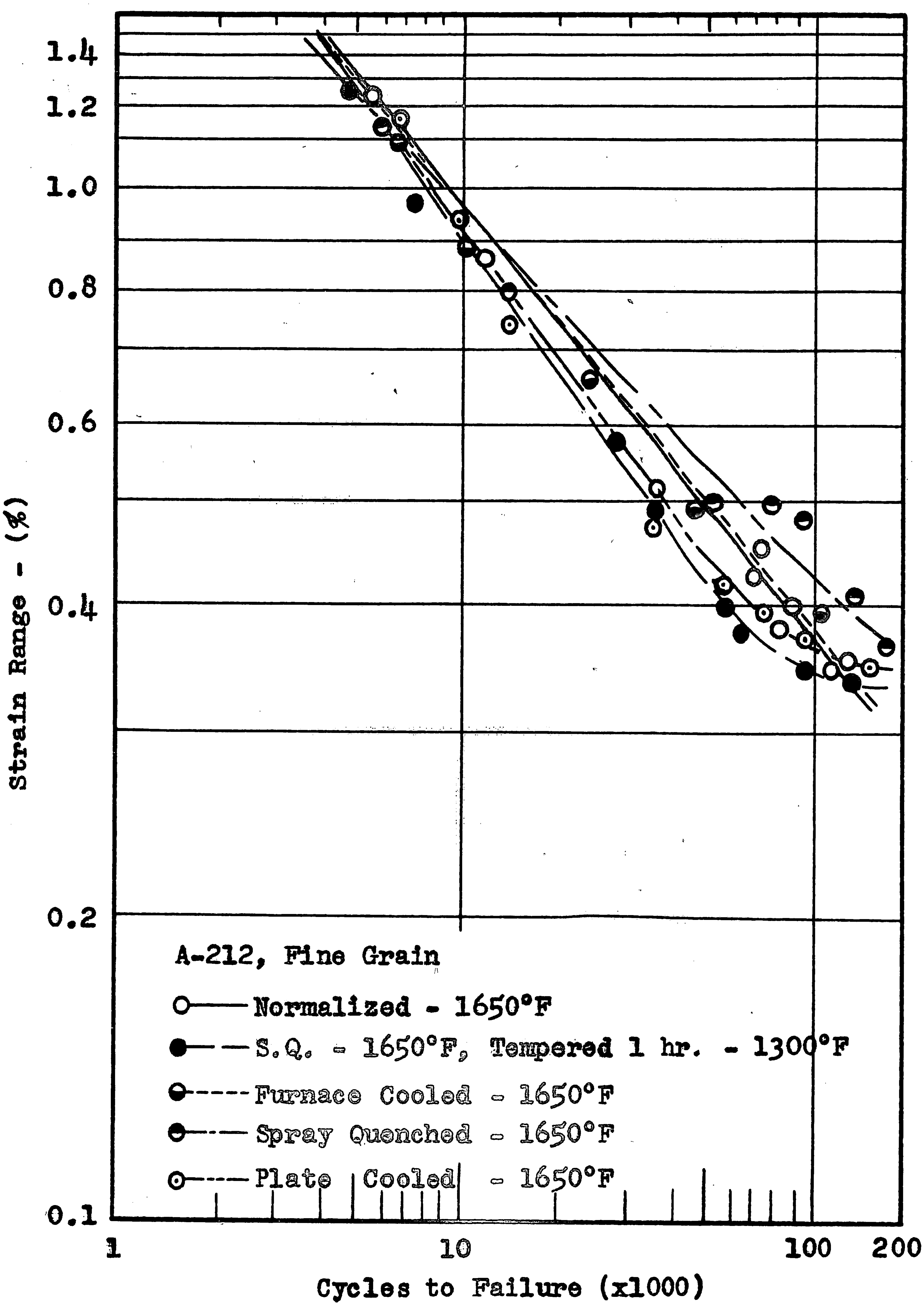


Figure 1. Lehigh Cantilever Fatigue Specimen.



1.4
1.2
1.0
0.8
0.6
0.4
0.2
0.1
Strain Range - (%)
A-212, Fine Grain
Normalized - 1650°F
S.Q. - 1650°F, Tempered 1 hr. - 1300°F
Furnace Cooled - 1650°F
Spray Quenched - 1650°F
Plate Cooled - 1650°F
1
10
100
200
Cycles to Failure (x1000)

Figure 2

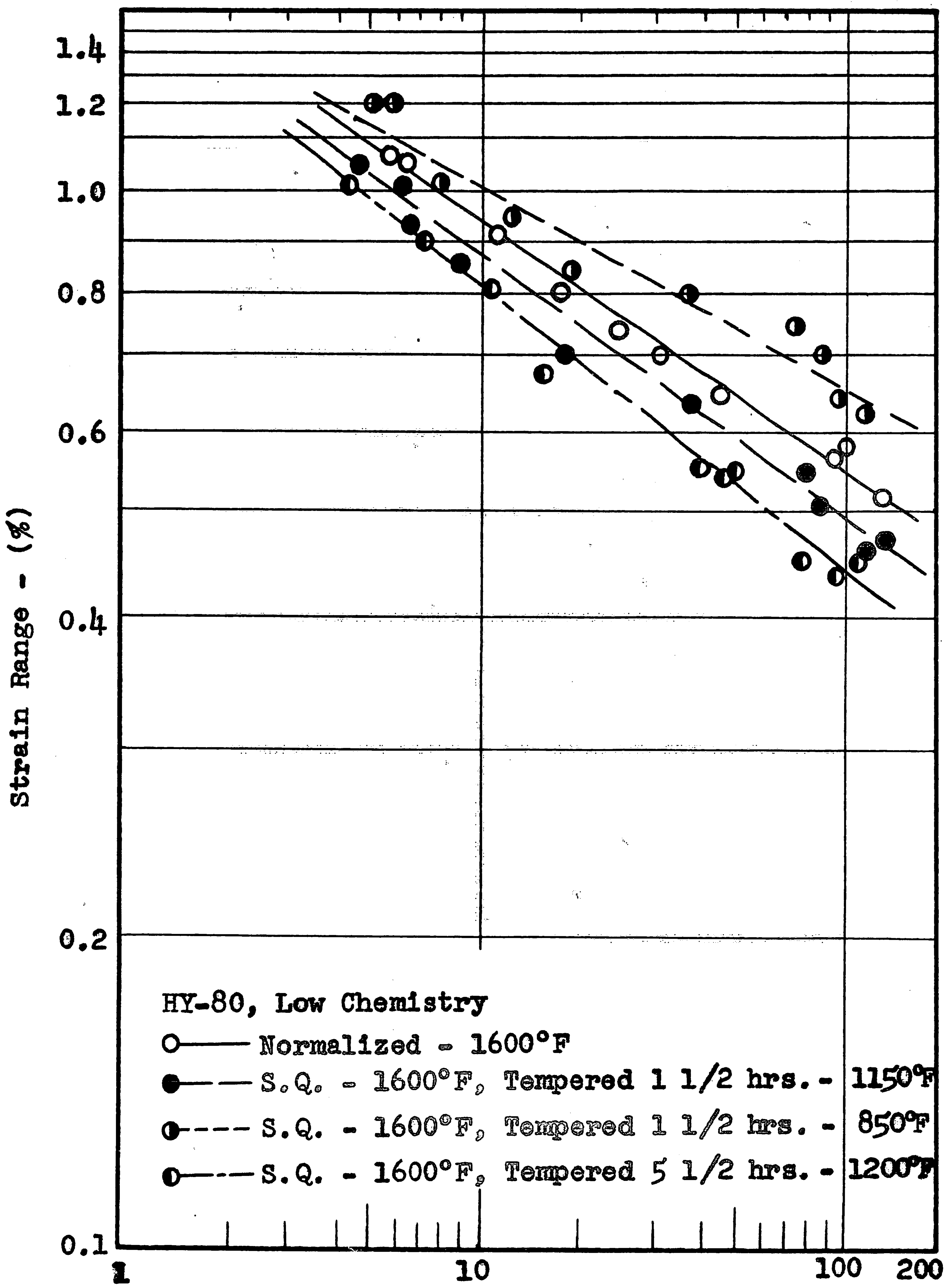
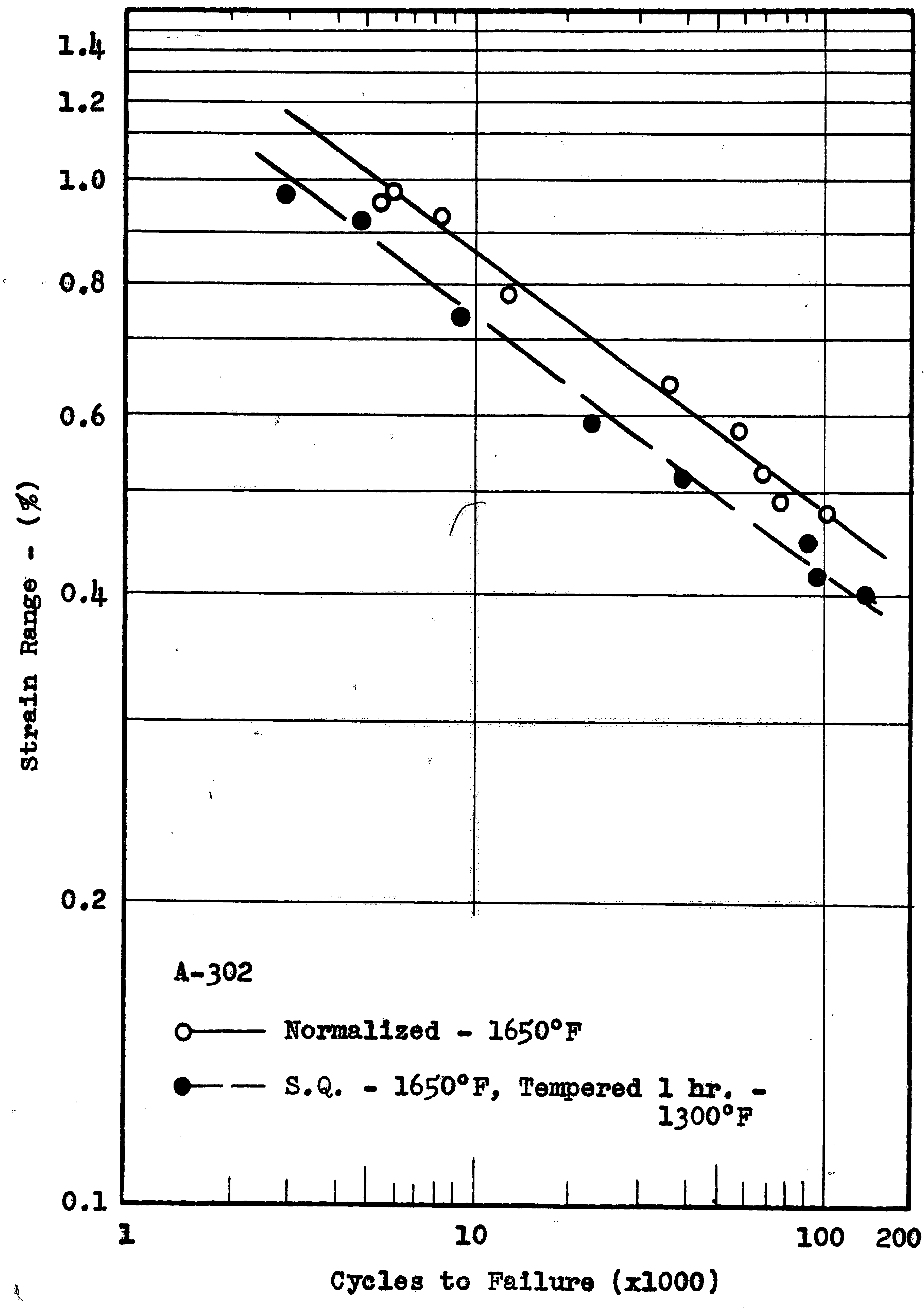

1.4
1.2
1.0
0.8
0.6
0.4
0.2
0.1
Strain Range - (%)
1
10
100
200
Cycles to Failure (x1000)
HY-80, Low Chemistry
Normalized - 1600°F
S.Q. - 1600°F, Tempered 1 1/2 hrs. - 1150°F
S.Q. - 1600°F, Tempered 1 1/2 hrs. - 850°F
S.Q. - 1600°F, Tempered 5 1/2 hrs. - 1200°F

Figure 3



1.4
1.2
1.0
0.8
0.6
0.4
0.2
0.1
Strain Range - (%)
A-302
Normalized - 1650°F
S.Q. - 1650°F, Tempered 1 hr. - 1300°F
1
10
100
200
Cycles to Failure (x1000)

Figure 4

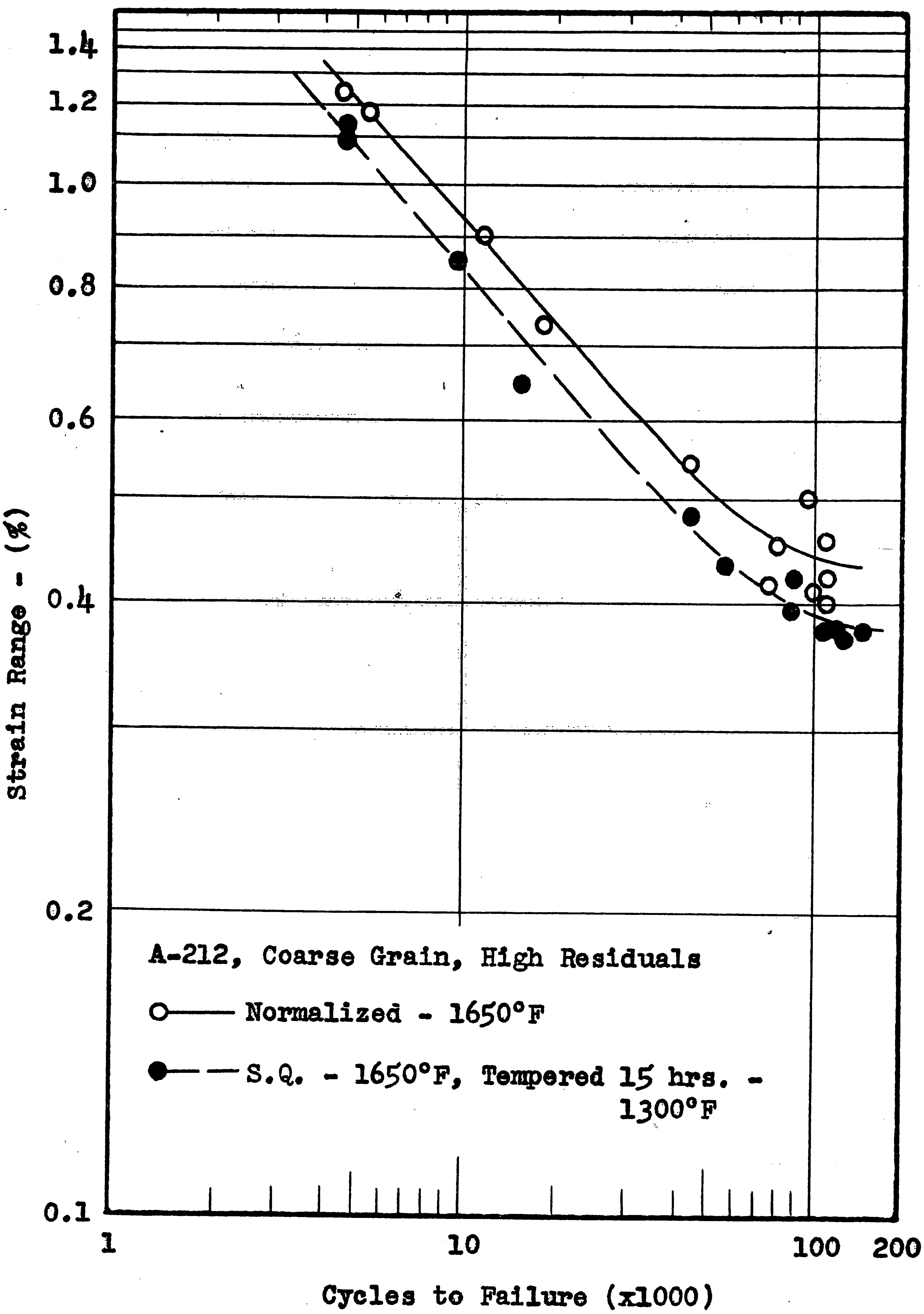

1.4
1.2
1.0
0.8
0.6
0.4
0.2
0.1
Strain Range - (%)
1
10
100
200
Cycles to Failure (x1000)
A-212, Coarse Grain, High Residuals
Normalized - 1650°F
S.Q. - 1650°F, Tempered 15 hrs. - 1300°F

Figure 5

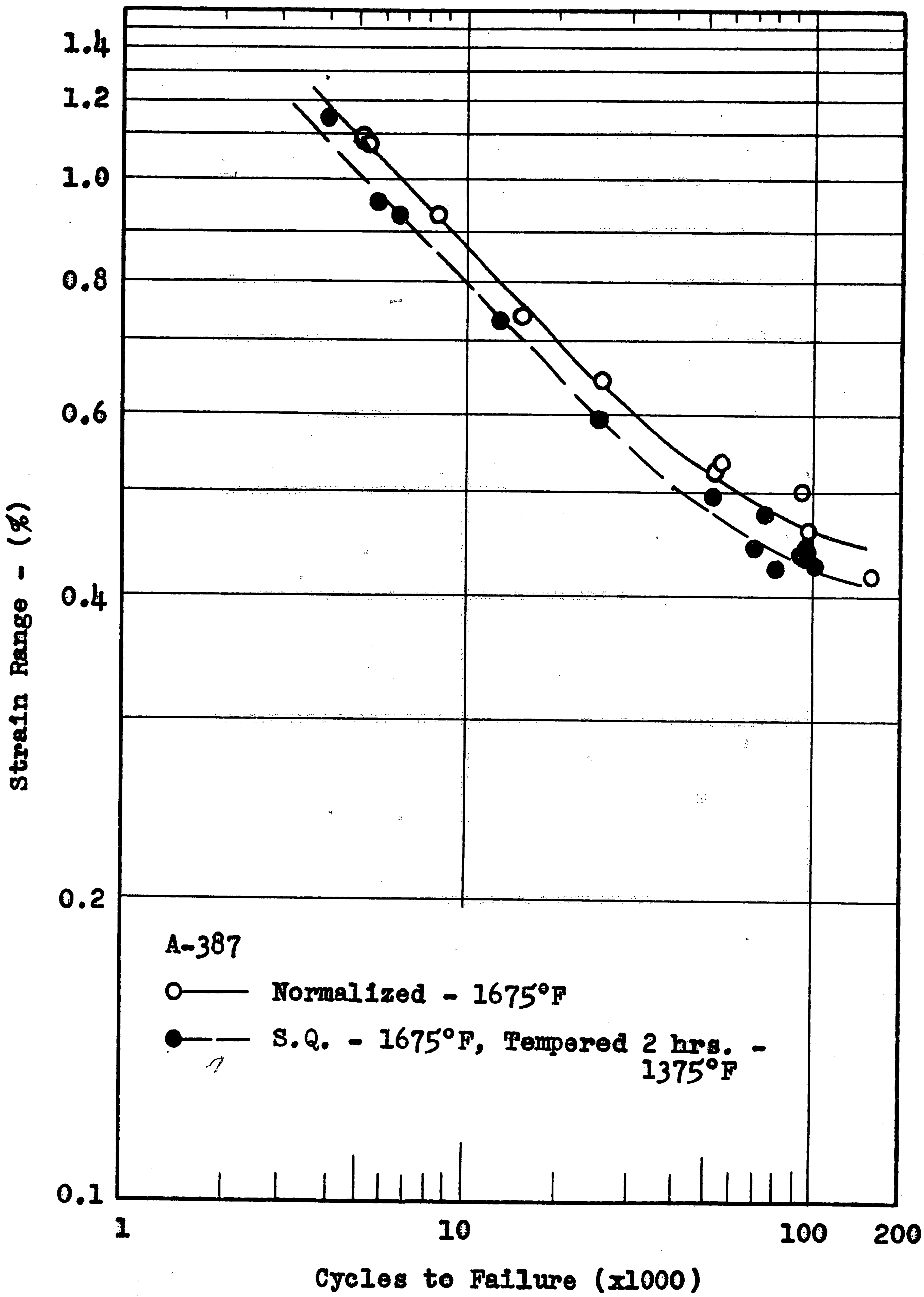
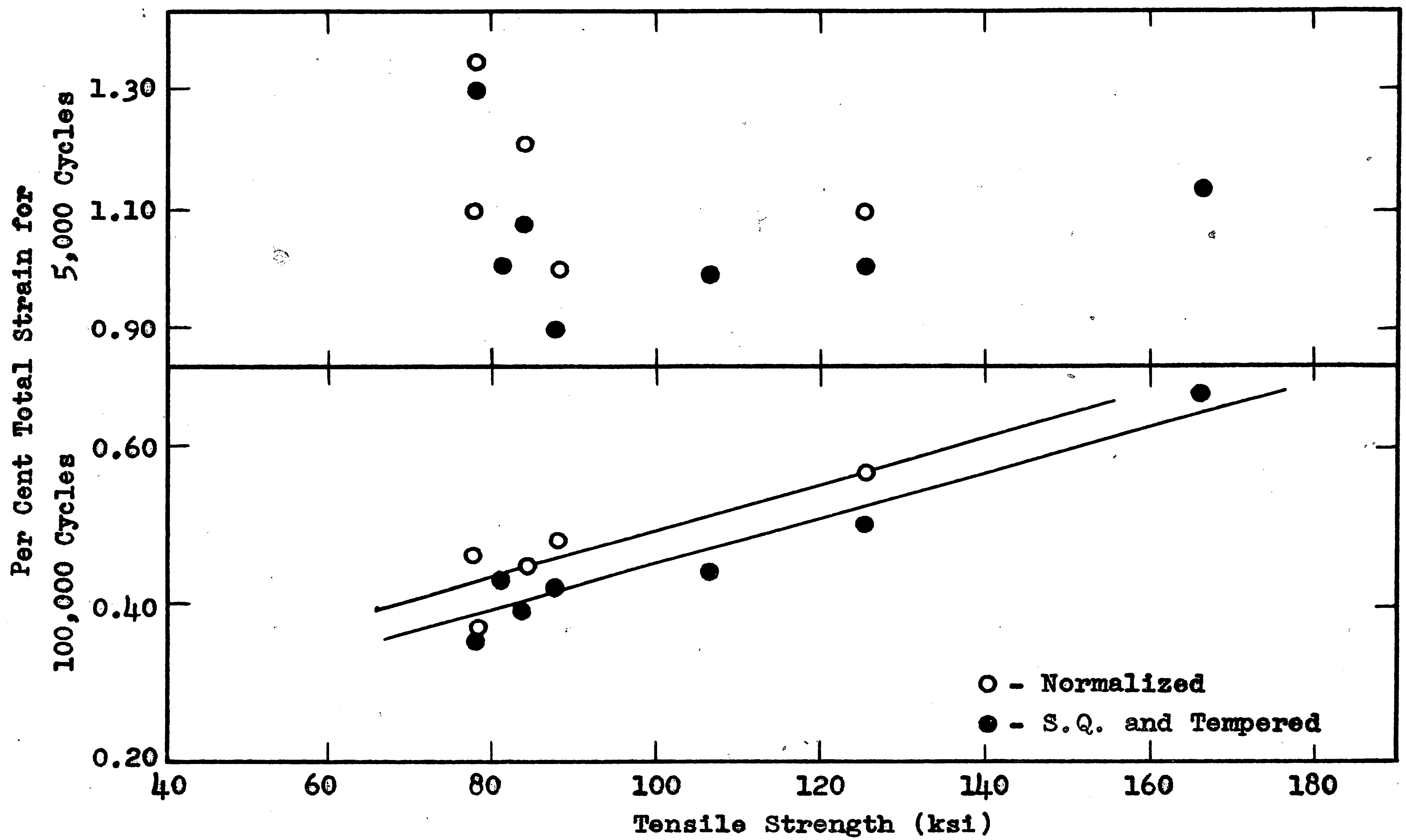

1.4
1.2
1.0
0.8
0.6
0.4
0.2
0.1
Strain Range - (%)
1
10
100
200
Cycles to Failure (x1000)
A-387
Normalized - 1675°F
S.Q. - 1675°F, Tempered 2 hrs. - 1375°F

Figure 6



Per Cent Total Strain for
5,000 Cycles
100,000 Cycles
1.30
1.10
0.90
0.60
0.40
0.20
40
60
80
100
120
140
160
180
Tensile Strength (ksi)
O - Normalized
● - S.Q. and Tempered

Figure 7

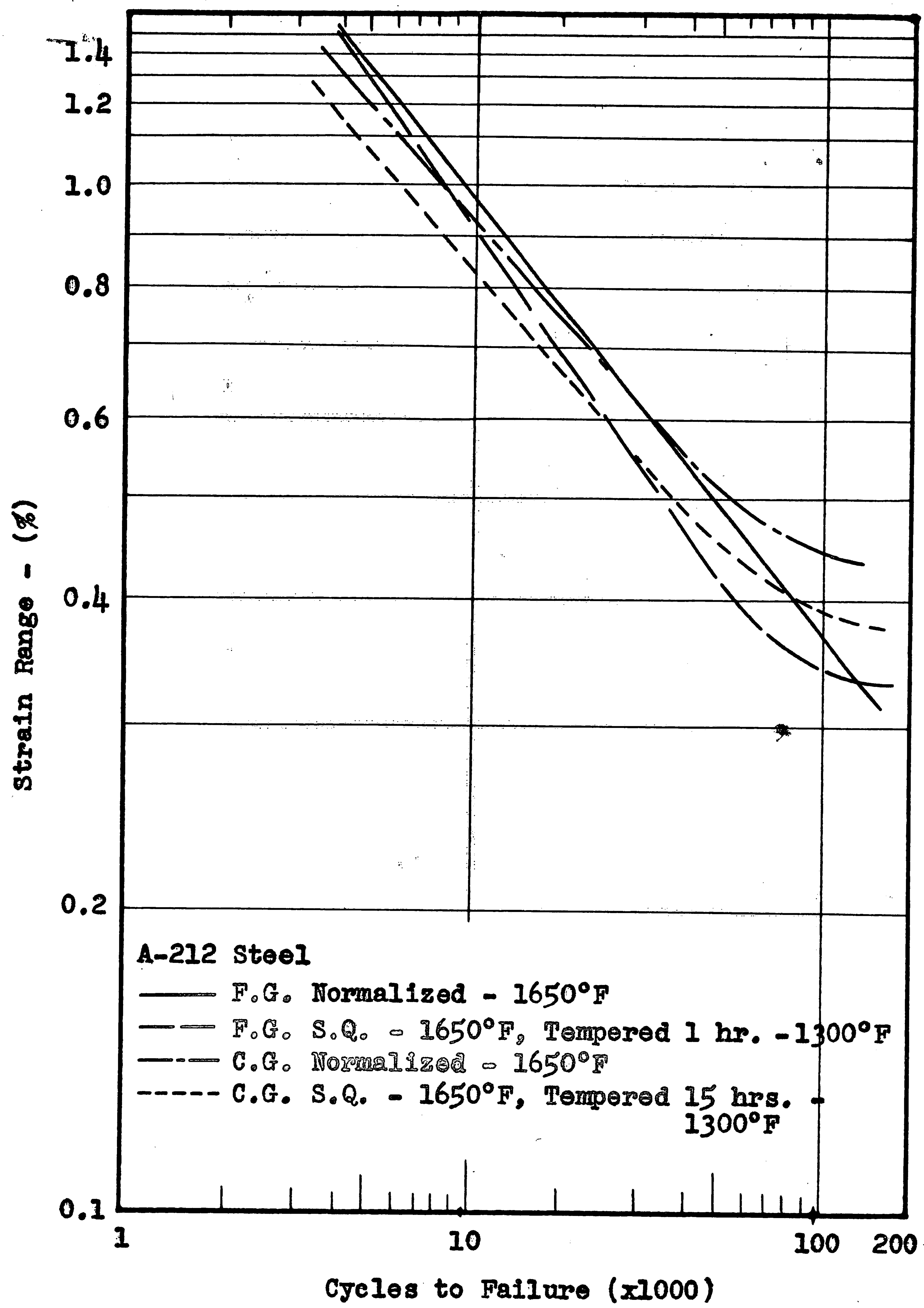

1.4
1.2
1.0
0.8
0.6
0.4
0.2
0.1
Strain Range - (%)
1
10
100
200
Cycles to Failure (x1000)
A-212 Steel
F.G. Normalized - 1650°F
F.G. S.Q. - 1650°F, Tempered 1 hr. -1300°F
C.G. Normalized - 1650°F
C.G. S.Q. - 1650°F, Tempered 15 hrs. - 1300°F

Figure 8

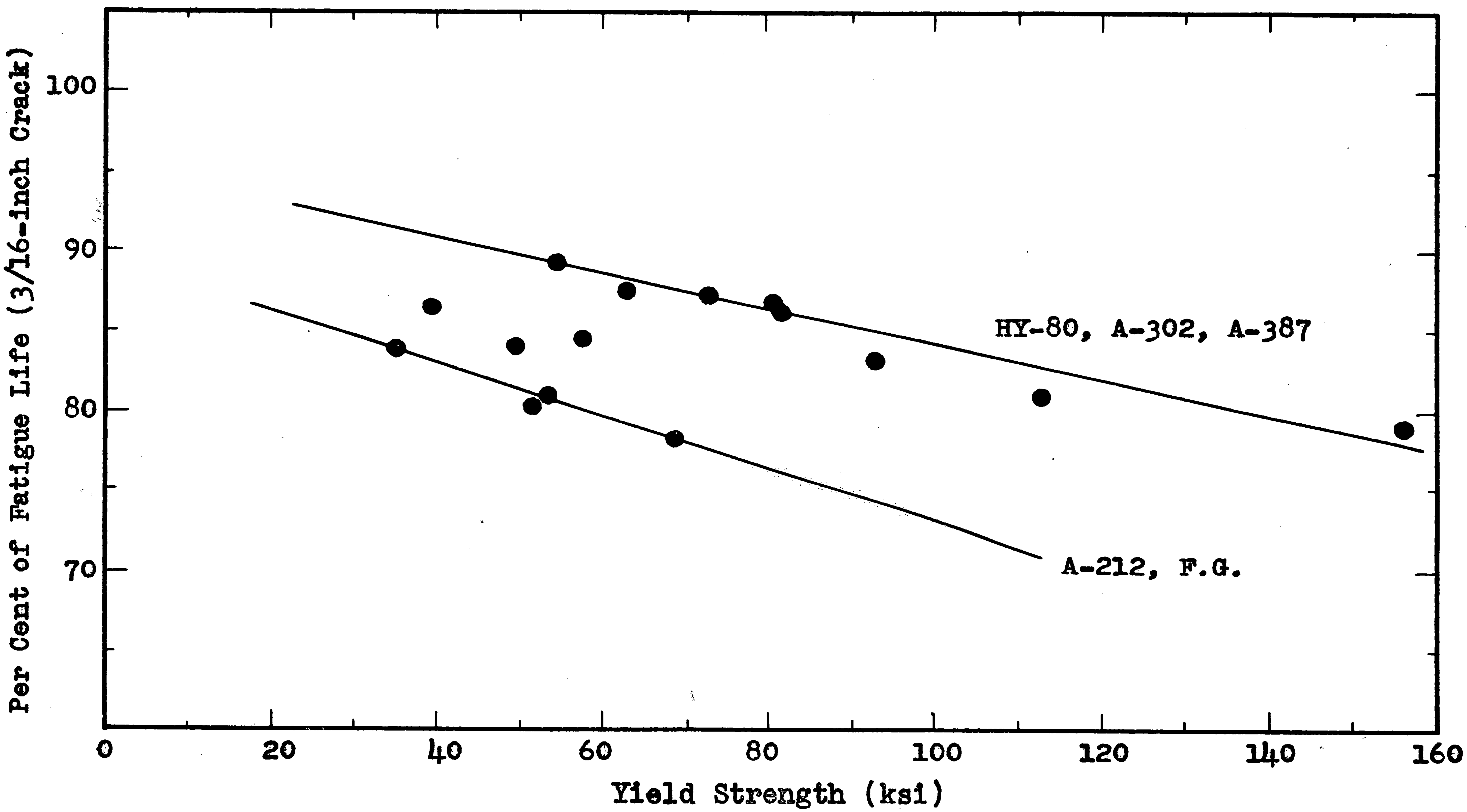
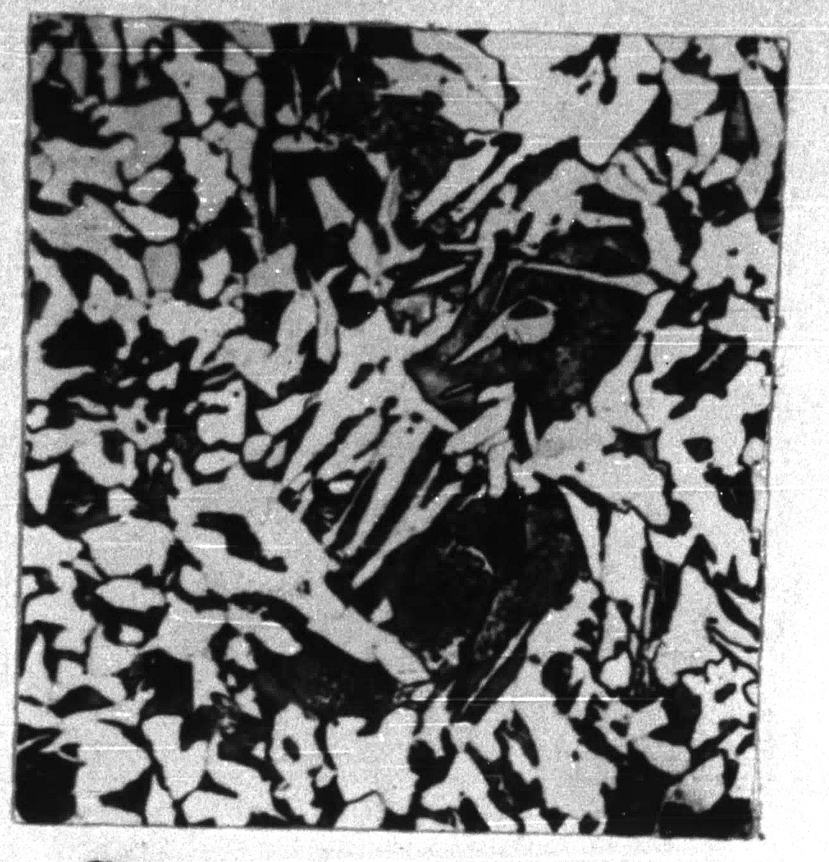

Per Cent of Fatigue Life (3/16-inch Crack)
100
90
80
70
0
20
40
60
80
100
120
140
160
HY-80, A-302, A-387
A-212, F.G.
Yield Strength (ksi)

Figure 9

APPENDIX A-1

MICROSTRUCTURES OF A-212 STEEL IN VARIOUS CONDITIONS OF HEAT TREATMENT



Spray Quenched
Cooling Rate
20°F/sec.

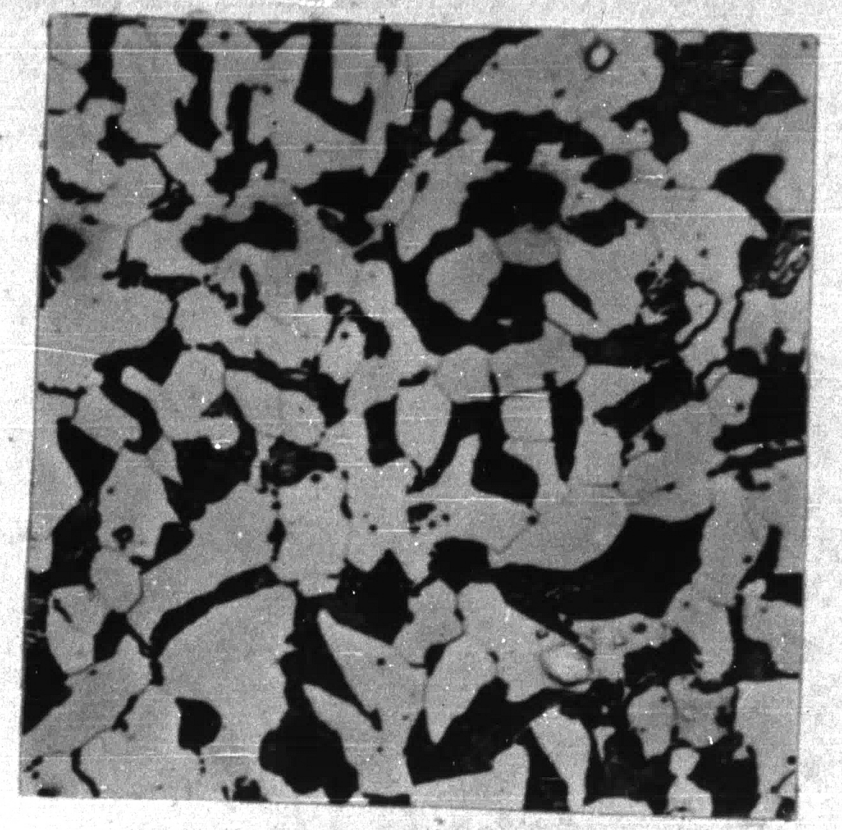
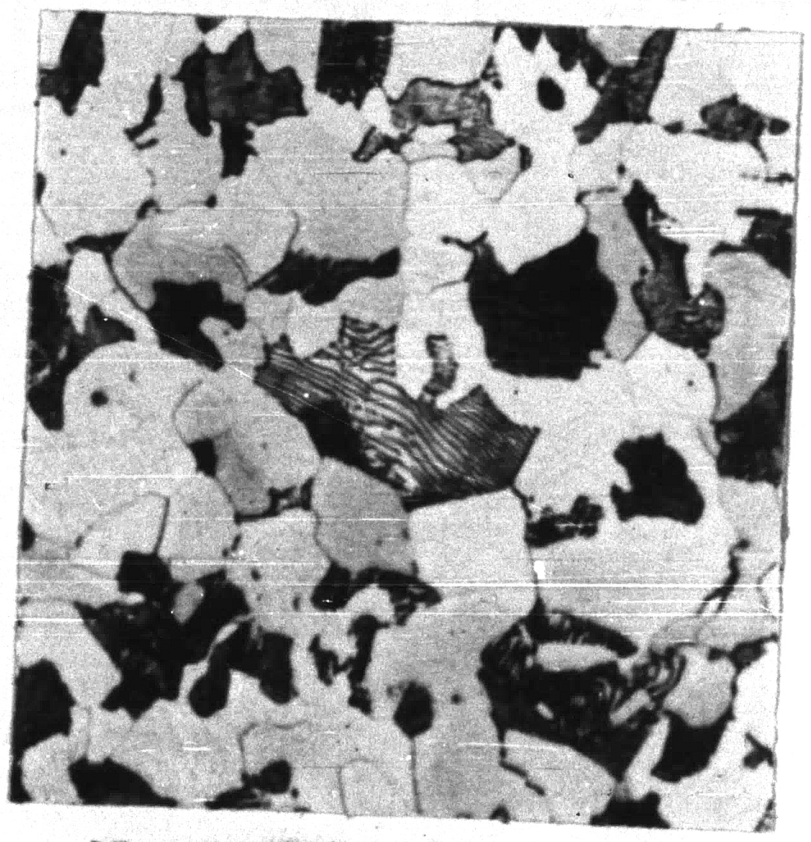
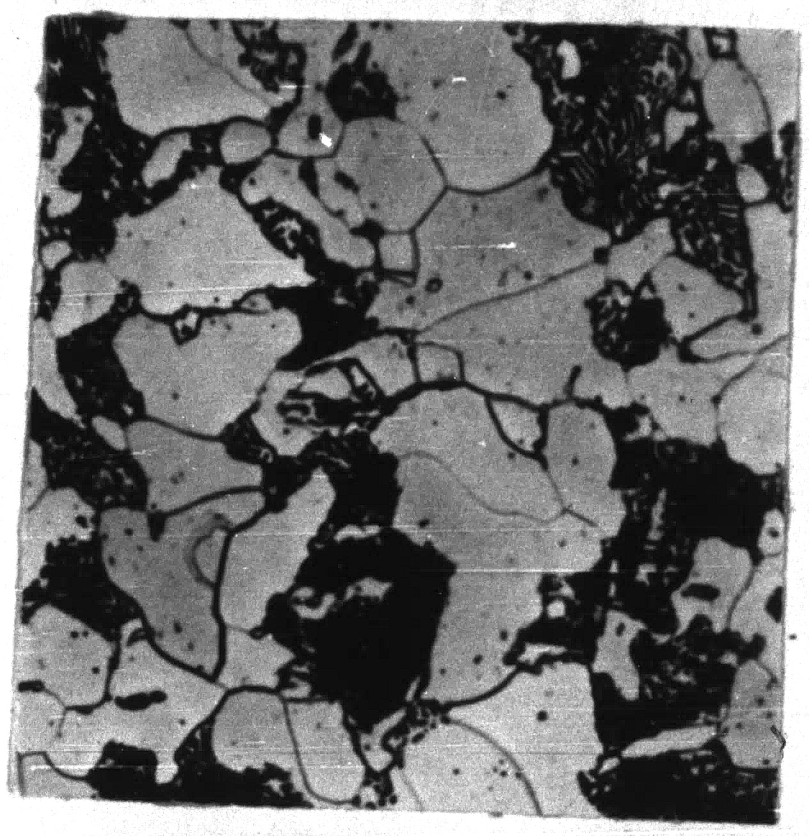


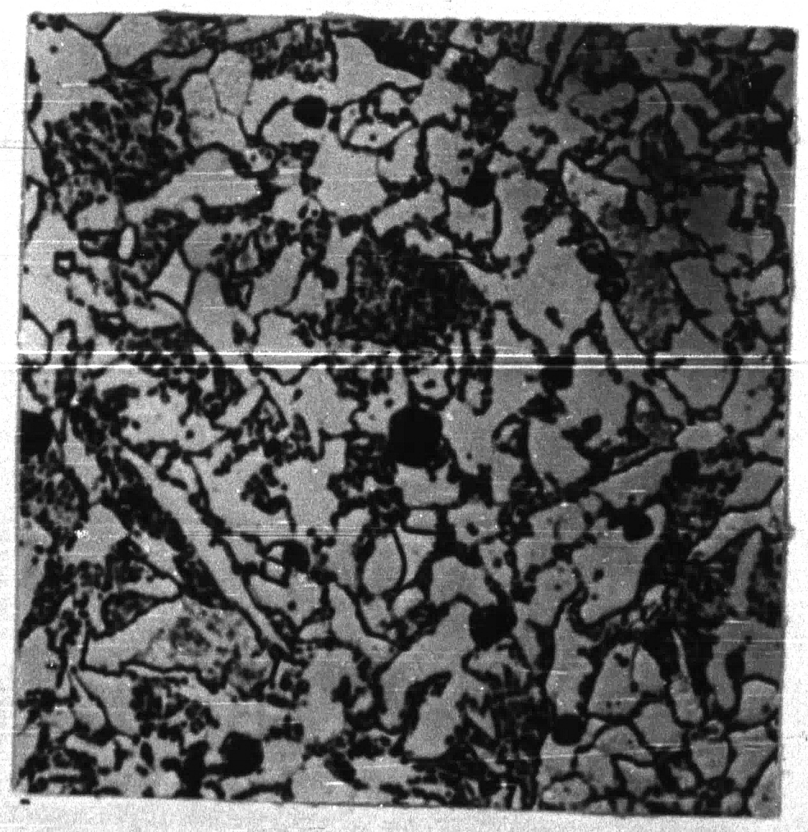
Plate Cooled
Cooling Rate
2.02°F/sec.



Normalized
Cooling Rate
1.01°F/sec.



Annealed
Cooling Rate
0.032°F/sec.



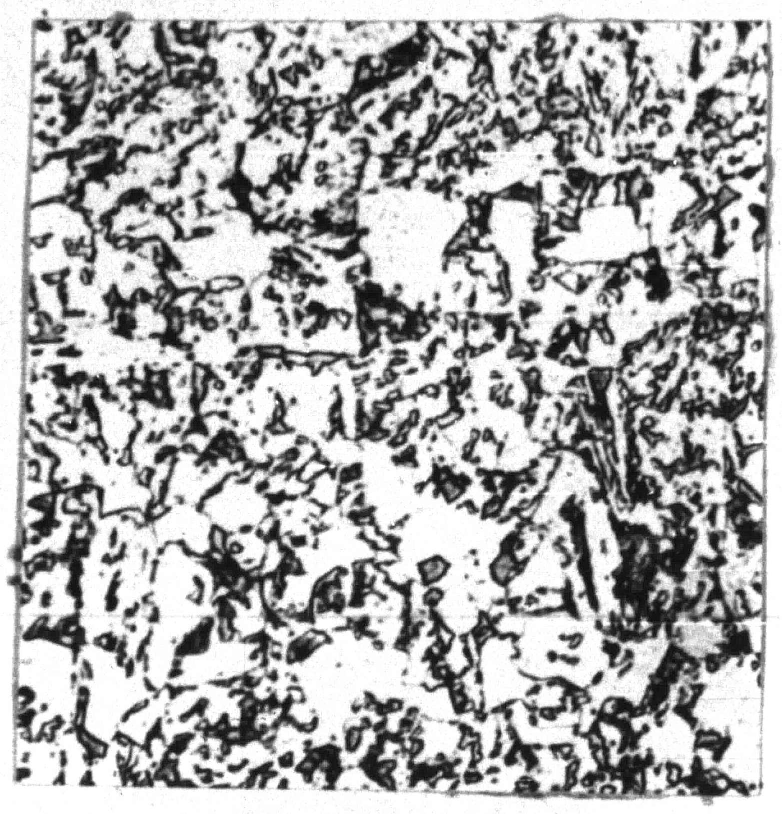
Spray Quenched and Tempered

500X

Nital Etch

APPENDIX A-2

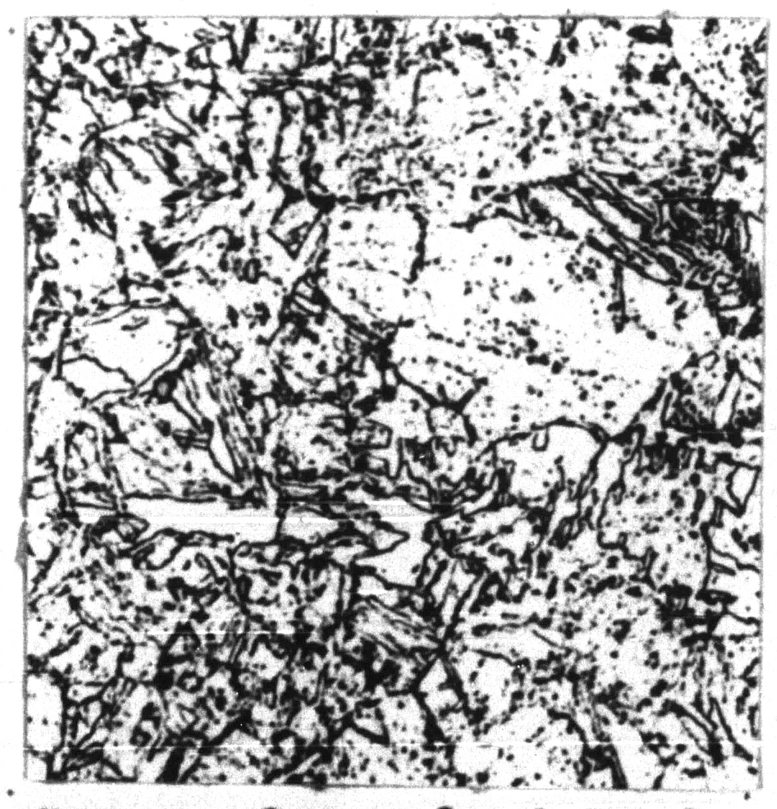
MICROSTRUCTURES OF HY-80 STEEL IN VARIOUS CONDITIONS OF HEAT TREATMENT



Normalized

Spray Quenched and Tempered 1 1/2 hrs. at 1150°F

Spray Quenched and Tempered 1 1/2 hrs. at 850°F



Spray Quenched and Tempered 5 1/2 hrs. at 1200°F

500X

Nital Etch

APPENDIX A-3

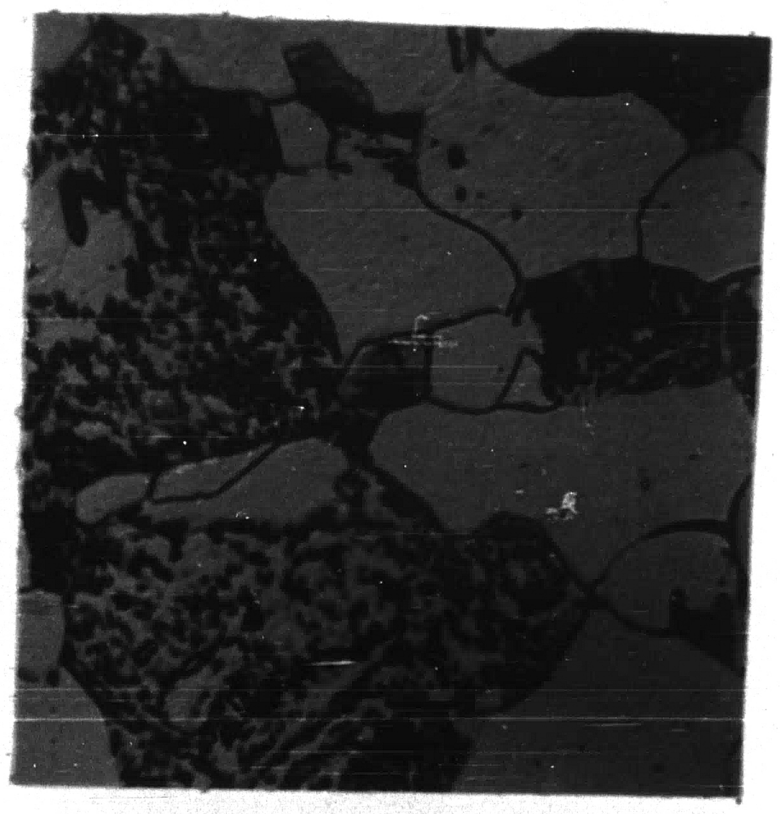
MICROSTRUCTURES OF A-212

COARSE GRAIN STEEL

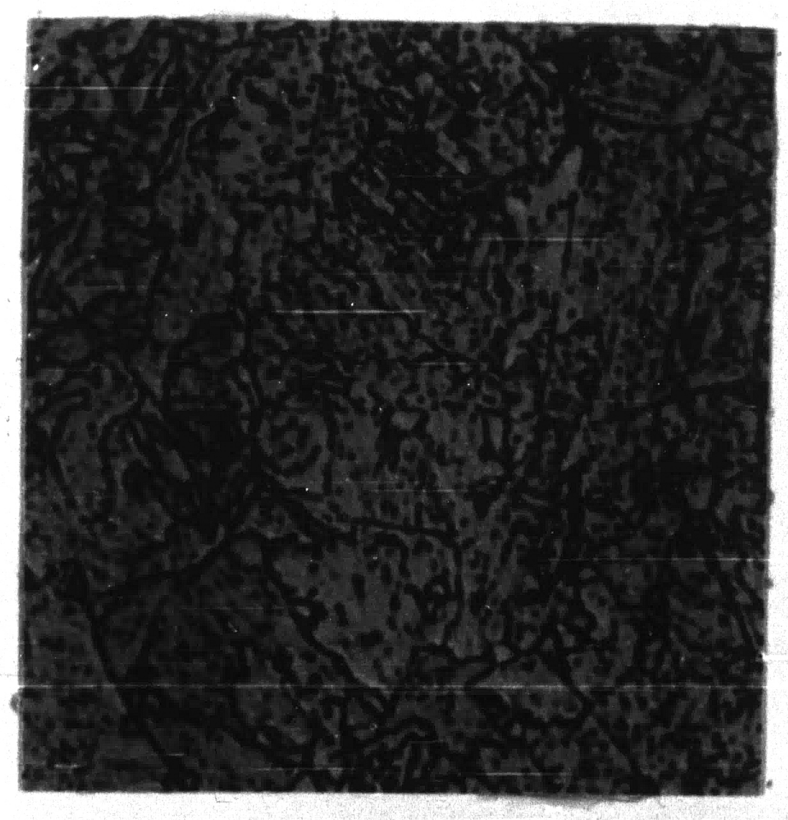
Normalized

Spray Quenched and Tempered

MICROSTRUCTURES OF A-387



Normalized



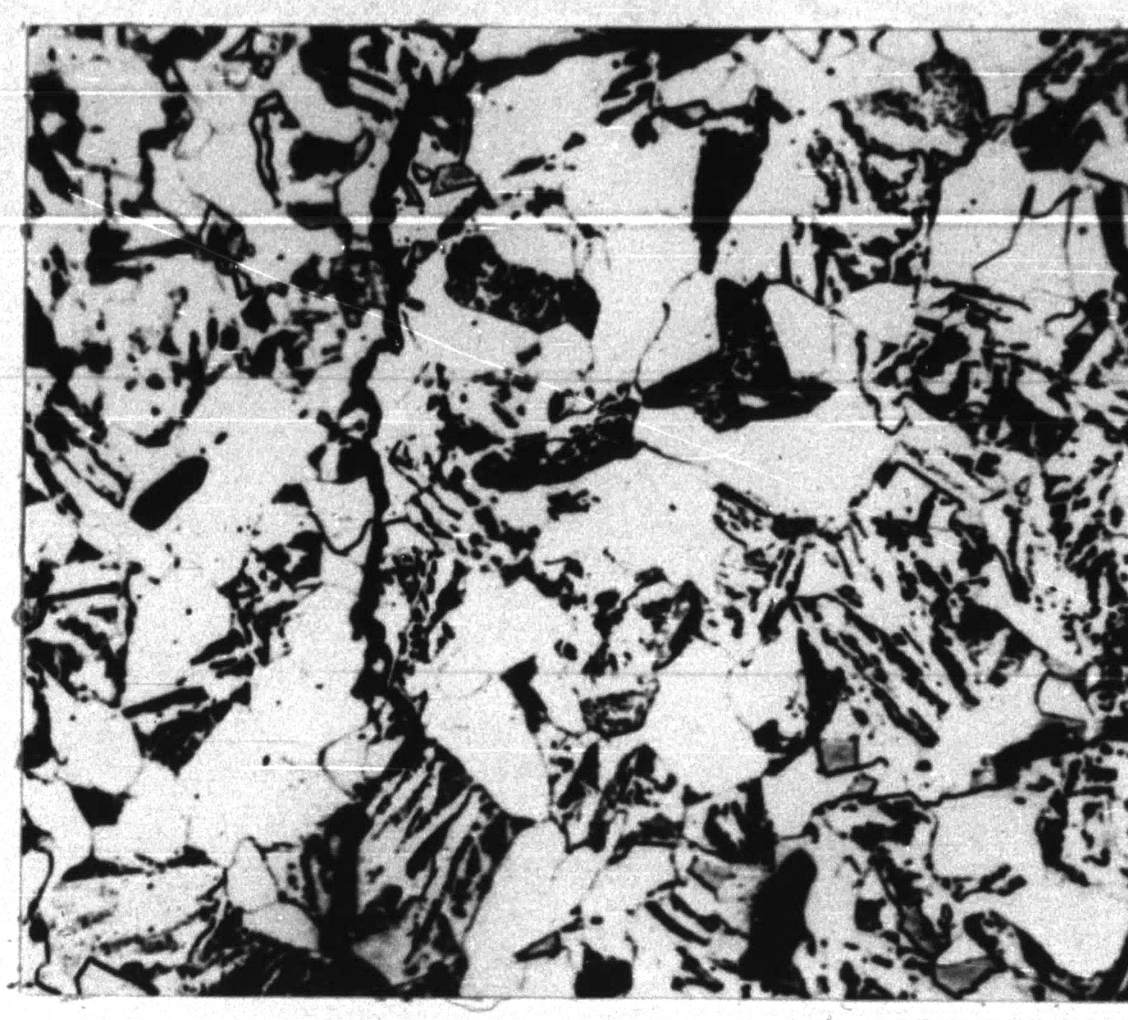
Spray Quenched and Tempered

500X

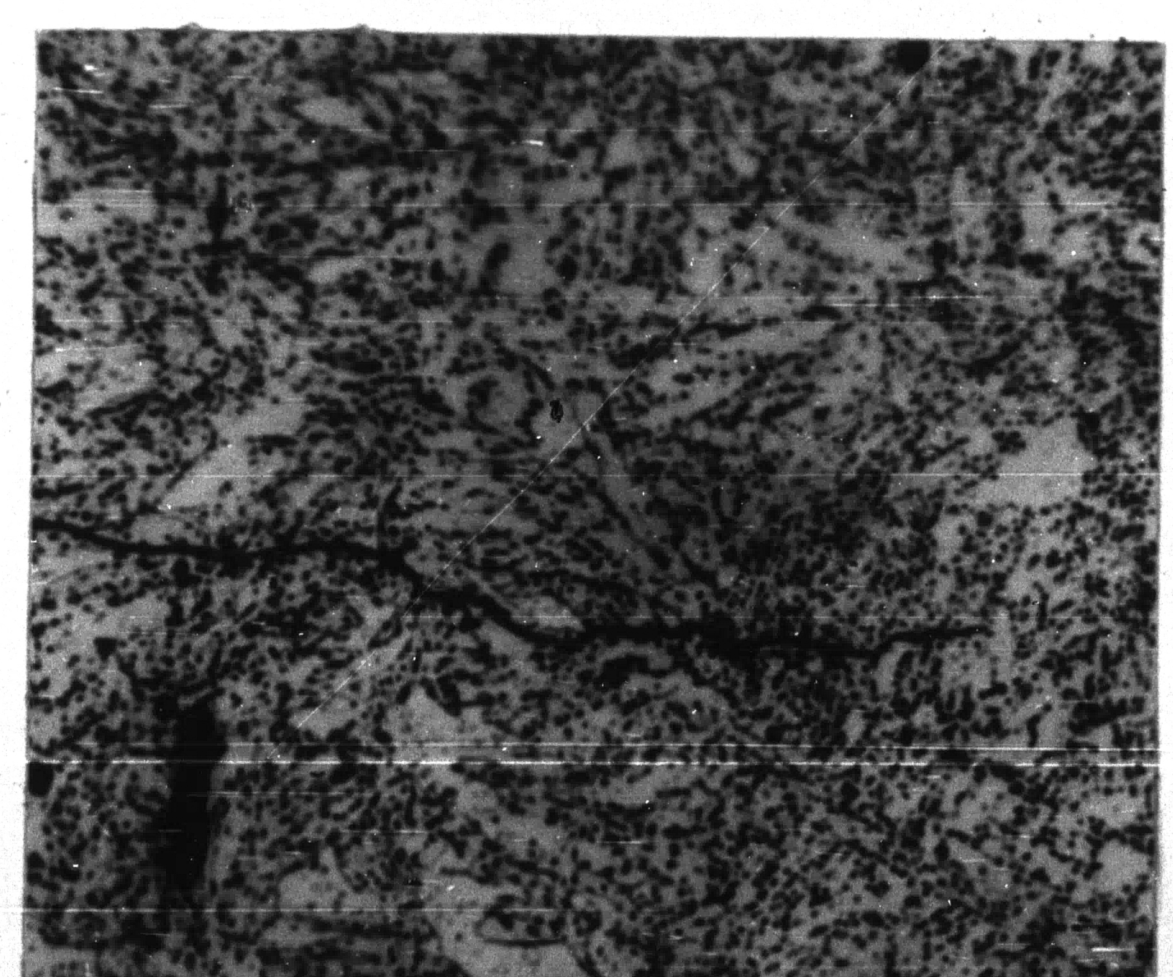
Nital Etch

APPENDIX A-4

BASE STRUCTURE AND TYPICAL FATIGUE CRACKS IN A-302 100,000 CYCLE LIFE



Normalized



Spray Quenched and Tempered

500X Nital Etch

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VITA

Robert Allen DePaul, son of Samuel and Emma M. DePaul, was born in Reading, Pennsylvania, on April 8, 1936. His elementary education was obtained in Reading, being graduated from Reading Senior High School in 1954. Following high school, Robert became associated with Textile Machine Works of Reading as a foundryman apprentice and attended Wyomissing Polytechnic Institute as part of the apprenticeship program. In 1956, he received a Certificate in Engineering Technology from Wyomissing and finished his apprentice training by being awarded his Journeyman's Papers in 1958. Robert was honored as an Outstanding Pennsylvania Apprentice in 1957 by the Pennsylvania Manufacturers' Association and was granted a three-year scholarship by Textile.

He entered Lehigh University in the fall of 1958 with advanced standing as a metallurgical engineering student. While an undergraduate at Lehigh, Robert was voted into Tau Beta Pi as a Junior and was awarded the Bradley Stoughton Student Award as a Senior. The writer was graduated with High Honors from Lehigh in 1961 receiving a B.S. in Metallurgical Engineering. He entered the Graduate School of Lehigh in the fall of 1961 where he was subsequently elected to the Society of the Sigma Xi. He is presently an instructor and research assistant in the Metallurgy Department of Lehigh University.